A local approach to cleavage fracture incorporating the measured statistics of microcracks

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A local approach to cleavage fracture incorporating the measured statistics of microcracks

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ABSTRACT. This work describes a local approach to cleavage fracture (LAF) incorporating the statistics of microcracks to characterize the cleavage fracture toughness distribution in structural steels. Fracture toughness testing conducted on standard compact tension C(T) specimens for a 22NiMoCr37 pressure vessel steel provides the cleavage fracture resistance data needed to determine the measured toughness distribution. Metallographic examination of etched surfaces for the tested steel also provides the distribution of carbides, which are assumed as the Griffith fracture-initiating particles, dispersed in the material from which the cleavage fracture toughness distribution is predicted. Overall, the analyses conducted in the present work show that LAFs incorporating the statistics of microcracks are a viable engineering procedure to describe the dependence of fracture toughness on temperature in the DBT region for ferritic steels.

KEYWORDS. Cleavage fracture, fracture toughness, local approach, carbide distribution, RPV steel

INTRODUCTION

There has been renewed interest in developing more rational micromechanics-based methodologies for cleavage fracture assessments, most often referred to as *local approaches to fracture* (LAFs) [1, 2, 3], which are essentially based on a probabilistic interpretation of the fracture process coupled with a local failure criteria describing unstable propagation of a critical microcrack. In particular, the seminal work of Beremin [1] builds upon the assumption that a power-law distribution describes the probability density function (*pdf*) for the size of Griffith-like microcracks that form immediately upon the onset of yielding ahead of a macroscopic crack to define the Weibull stress (σ_w) as a probabilistic fracture parameter. Within the Beremin model, σ_w is shown to follow a two-parameter Weibull distribution defined by the Weibull modulus, m, and the scale parameter, σ_u, in which the key parameter m is assumed to represent a material property [1, 2, 3] thereby providing the necessary framework for stress-controlled cleavage fracture assessments.

While LAF methodologies represent a major advancement in current fracture assessment procedures to analyze the significance of defects and material degradation, difficulties still persist in engineering applications of the probabilistic model based on the Weibull stress concept that can have important implications for more accurate structural integrity assessments. Specifically, a major point of concern lies in the rather strong sensitivity of fracture predictions to the calibrated Weibull modulus, which is most often obtained based on limited experimental data [4] - typical values for parameter m range from 10 to 20 for structural ferritic materials, including common pressure vessel steels. Here, since σ_w scales with $\int \sigma_1^m$ performed over the near-tip fracture process zone (see full details in Ruggieri and Dodds (R&D) [4]), larger m values assign a greater weight factor to stresses at locations very near the crack front but, at the same time, reducing the potential contributions of near-tip material volume on the fracture probability. Hence, the resulting analysis based on the interpretation of σ_w as a macroscopic crack driving force may not uncover the controlling microfeatures for cleavage fracture, such as the actual fraction of eligible Griffith-like microcracks nucleated from the brittle particles effectively controlling cleavage fracture, thereby making the Weibull stress methodology insuficiently robust to serve as a more rigorous framework for cleavage fracture assessments.

Motivated by these observations, this work describes a local approach to cleavage fracture incorporating the statistics of microcracks to characterize the cleavage fracture toughness distribution in structural steels. One purpose of this study is to explore a theoretical framework consistent with what exists for probabilistic modeling of cleavage fracture to develop a failure probability model derived entirely from the measured carbide distribution dispersed into the ferritic matrix. Another is to explore application of the probabilistic model to describe the temperature dependence of J_c-values in the ductile-to-brittle (DBT) region for a nuclear pressure vessel steel. Fracture toughness testing conducted on standard compact tension C(T) specimens for a 22NiMoCr37 pressure vessel steel, known as the "Euro" Material A, provides the cleavage fracture resistance data needed to determine the experimentally measured J_c-distribution. Metallographic examination of etched surfaces for the Euro A steel also provides the distribution of carbides, which are assumed as the Griffith fracture-initiating particles, dispersed in the material from which the cleavage fracture toughness distribution is predicted. Overall, the analyses conducted in the present work show that LAFs incorporating the statistics of microcracks are a viable engineering procedure to describe the dependence of fracture toughness on temperature in the DBT region for ferritic steels.

CLEAVAGE FRACTURE MODELING INCORPORATING STATISTICS OF MICROCRACKS

Consider the near-tip fracture process zone (FPZ) ahead of a stationary macroscopic crack lying in a material containing randomly oriented microcracks, uniformly distributed in location. Further consider that the near-tip FPZ is idealized as consisting of a large number of statistically independent, uniformly stressed, small volume elements, denoted δV, and which are subjected to the principal stress, σ_1, and associated effective plastic strain, ε_p, as illustrated in Fig. 1(a). Now let N_c be the number of microcracks nucleated from fractured carbides in the small volume, δV, given by

$$N_c = \Psi_c \rho_d \delta V \tag{1}$$

where ρ_d is the (average) density of carbides in the FPZ material and Ψ_c represents the fraction of fractured carbides which are assumed as the Griffith-like microcracks that are eligible to propagate unstably $(0 \le \Psi_c \le 1)$.

The specific micromechanism of transgranular cleavage allows assuming that failure of each small volume element occurs when the size of a random microcrack contained in δV exceeds a critical size, *i.e.*, $a \ge a_c$. Thus, using weakest link arguments in connection with the assumption that cleavage fracture is governed by the failure of a small volume element, δV, the failure probability for the cracked body, P_f, is expressed as

$$P_f = 1 - \exp\left\{ \int_{V_{FPZ}} \ln\left[\int_0^{a_c} g(a)da \right]^{N_c} dV_{FPZ} \right\} \tag{2}$$

in which $g(a)da$ defines the probability of finding a microcrack having size between a and $a + da$ in the small volume, and it is understood that the integral is performed over the volume of the near-tip fracture process zone, V_{FPZ}.

In the above, the critical microcrack size, a_c, follows from a modified form of the Griffith criterion [5] expressed as

$$a_c = \frac{2E\gamma_f}{\pi(1-\nu^2)\sigma_1^2} \tag{3}$$

where E is the Young's modulus, ν denotes the Poisson's ratio and σ_1 represents the maximum principal stress acting normal to the microcrack plane. Here, $\gamma_f = \gamma_s + \gamma_p$ defines the effective fracture energy to propagate the microcrack in which γ_s is the (elastic) surface energy and γ_p is the temperature dependent, plastic work per unit area of surface created.

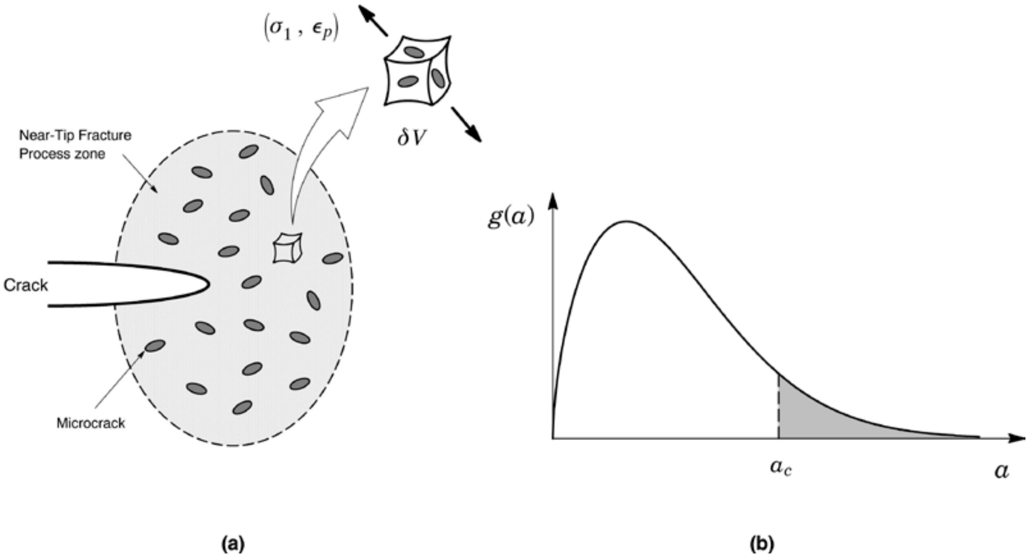
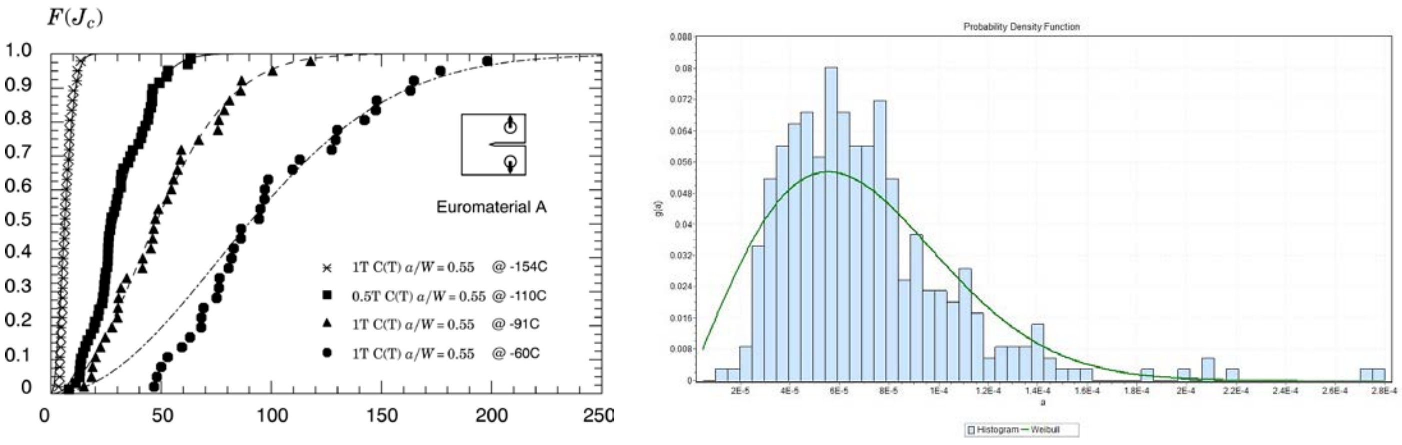


Figure 1: (a) Near-tip fracture process zone ahead a macroscopic crack containing randomly distributed flaws b) Schematic of probability density function (*pdf*) to describe the microcrack size distribution.

FRACTURE TOUGHNESS TESTS RESULTS AND MATERIAL CHARACTERIZATION

Extensive fracture toughness tests were performed on a nuclear reactor pressure vessel (RPV) class steel DIN 22NiMoCr37 similar to ASTM A508 Cl.3 steel and widely referred to as "Euro" Material A. Conducted as part of the "Measurement and Testing Programme of the European Commission", the testing program focused primarily on developing an experimental fracture toughness data base for validation of the Master Curve methodology, including an experimental investigation of specimen geometry and temperature effects on fracture toughness in ferritic materials. Heerens and Hellmann (H&H) [6] provide a detailed description of the Euro fracture toughness dataset. Here, we limit attention to selected fracture toughness distributions measured at four test temperatures: $T = -154^\circ\text{C}, -110^\circ\text{C}, -91^\circ\text{C}$ and -60°C. H&H [6] also provide the mechanical properties for the tested material for these test temperatures.

The cumulative Weibull distributions of the measured J_c-values are displayed in Fig. 2(a) in which the solid symbols in the plots represent the experimentally measured fracture toughness. The fracture toughness tests were performed on conventional, plane-sided compact tension C(T) specimens with $a/W \approx 0.55$. The fracture mechanics tests include: (1) 0.5T C(T) specimens ($B = 12.5\,\text{mm}$) tested at $T = -110^\circ\text{C}$; (2) 1T C(T) specimens ($B = 25\,\text{mm}$) tested at $T = -154^\circ\text{C}$, -91°C and $T = -60^\circ\text{C}$. Here, a is the crack size and W denotes the specimen width. In this plot, the cumulative probability, $F(J_c)$, is derived by simply ranking the J_c-values in ascending order and using the median rank position defined in terms of $F(J_c) = (k-0.3)/(N+0.4)$, where k denotes the rank number and N defines the total number of experimental toughness values. The fitting curves to the experimental data shown in this figure describe the three-parameter Weibull distribution for J_c-values with a fixed value of $\alpha = 2$ as the Weibull modulus and a threshold J-value corresponding to a $K_{\min} = 20\,\text{MPa}\sqrt{\text{m}}$.



(a)

(b)

Figure 2: (a) Cumulative Weibull distribution of the measured J_c-values for the "Euro" Material A at test temperatures: $T = -154^\circ\mathrm{C}, -110^\circ\mathrm{C}, -91^\circ\mathrm{C}$ and $-60^\circ\mathrm{C}$. (b) Histogram of the particle size distribution measured by Ortner and Duff [8], including the fitted two-parameter Weibull *pdf*.

FINITE ELEMENT PROCEDURES

Nonlinear finite element analyses are performed for 3-D models of the tested C(T) specimens with $a/W \approx 0.55$ described previously. Figure 3 shows the typical finite element model utilized in the analyses of the 1T C(T) specimen with $B = 25\,\mathrm{mm}$. With minor differences, the numerical model for 0.5T C(T) specimen with $B = 12.5\,\mathrm{mm}$ has very similar features. A conventional mesh configuration having a focused ring of elements surrounding the crack front is used with a small key-hole at the crack tip with radius $\rho_0 = 0.0025\,\mathrm{mm}$. Symmetry conditions enable analyses using one-quarter of the 3-D models with appropriate constraints imposed on the symmetry planes. The quarter-symmetric, 3-D model for this specimen has 25 variable thickness layers and approximately 40,000 nodes and 36,000 8-node, 3-D elements.

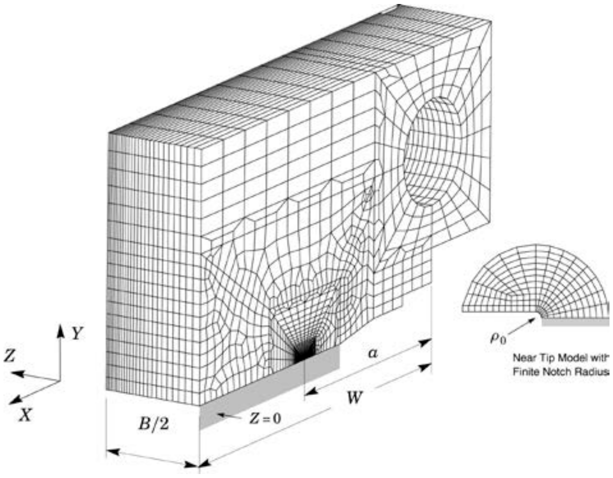


Figure 3: 3-D FE model of the 1T C(T) specimen with $a/W \approx 0.55$ and $B = 25\,\mathrm{mm}$.

The finite element code WARP3D [7] provides the numerical solutions for the 3-D analyses reported here. The finite element analyses utilize an elastic-plastic constitutive model with J_2 flow theory and conventional Mises plasticity in large geometry change (LGC) setting incorporating a simple power-hardening model to characterize the uniaxial true stress *vs*. logarithmic strain with the hardening exponent estimated from the flow properties given by H&H [6].

PARAMETER CALIBRATION AND FRACTURE TOUGHNESS PREDICTIONS

Calibration of the probabilistic model follows from matching the predicted toughness distribution with the experimentally measured J_c-values for the 1T C(T) specimen tested at $T = -154^\circ\text{C}$. This specimen arguably provides the lowest levels of crack-tip plastic deformation and thus minimizes the dependence of the critical microcrack size, a_c, on the temperature dependent, plastic work γ_p. Moreover, based on the carbide size measurements for the Euro Material A conducted by Ortner and Duff [8], we adopt a two-parameter Weibull distribution to describe the *pdf* for the carbide size, $g(a)$, appearing in Eq. (2) and expressed as

$$g(a) = \frac{\alpha_c}{\beta_c}\left(\frac{a}{\beta_c}\right)^{\alpha_c - 1} \exp\left[-\left(\frac{a}{\beta_c}\right)^{\alpha_c}\right] \tag{4}$$

in which a fitting procedure yields $\alpha_c = 1.996$ and $\beta_c = 7.876 \cdot 10^{-5}\ \text{mm}$. Figure 2(b) shows the histogram of the particle size distribution corresponding to an average density of carbides, $\rho_d = 7.6 \cdot 10^8\ \text{mm}^{-3}$ measured by Ortner and Duff [8] which also includes the fitted two-parameter Weibull distribution.

Figure 4(a) displays the predicted toughness distribution for the 1T C(T) specimen tested at $T = -154^\circ\text{C}$ derived from solving Eq. (2) in connection with Eq. (1) and Eq. (3) at each load step with the fraction of fractured carbides, Ψ_c, described by a Pareto type function [9] in the form

$$\Psi_c = 1 - \left(\frac{\varepsilon_{ys}}{\varepsilon_p}\right)^2 \tag{5}$$

where $\varepsilon_{ys} = \sigma_{ys}/E$ with σ_{ys} representing the yield stress at the test temperature. For the Euro Material A at $T = -154^\circ\text{C}$, adopting an effective fracture energy of $\gamma_f = 6\ \text{J/m}^2$ provides good agreement between the measured and predicted toughness distributions.

To arrive at the toughness distributions for other test temperatures, we adopt the following procedure. First, we determine the predicted toughness distribution at $T = -60^\circ\text{C}$ with $\gamma_f = 7.3\ \text{J/m}^2$ as illustrated in Fig. 4(b). Next, by adopting the surface energy of iron as $\gamma_s = 2\ \text{J/m}^2$ and following similar procedure outlined in Wallin et al. [10], we determine the temperature dependence of γ_p and, consequently, γ_f in the form

$$\gamma_f = 2 + 6.35 \cdot \exp(0.003T) \tag{6}$$

where T is the test temperature in degrees Celsius. Figure 4(b) also shows the predicted toughness distributions for the C(T) specimens tested at $T = -110^\circ\text{C}$ and -91°C in which general good agreement with the measured data is observed.

CONCLUDING REMARKS

This work describes an exploratory application of a local approach to cleavage fracture incorporating the measured distribution of carbides and brittle particles. The analyses conducted in the present work show that predictions of the temperature dependence of J_c-values are in good agreement with the measured toughness distributions. However, calibration of the key parameters controlling cleavage fracture still remains rather complex for robust applications of the probabilistic model in routine structural integrity assessments. In particular, a clear and simple estimation procedure for the temperature dependent, plastic work parameter, γ_p, is still considered an open issue and work along this line of investigation is in progress. Overall, the present analyses show that LAFs incorporating the statistics of microcracks appear a viable engineering procedure to describe the dependence of fracture toughness on temperature in the DBT region for ferritic steels.

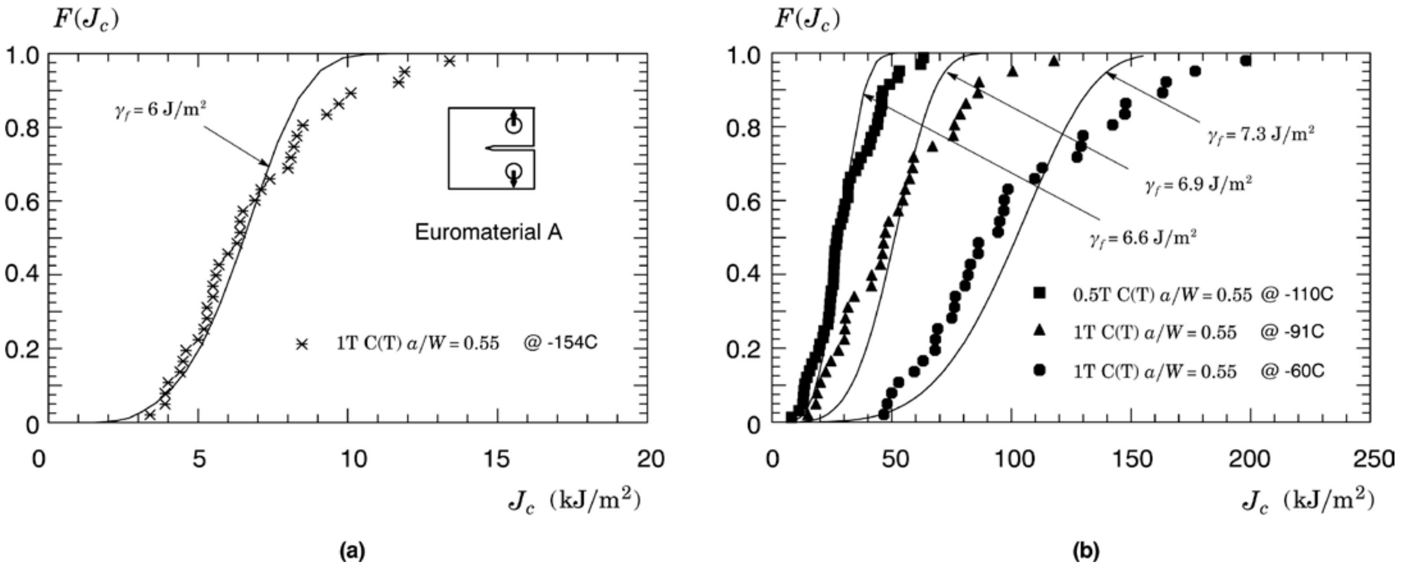


Figure 4: (a) Predicted toughness distribution for the 1T C(T) specimen tested at $T = -154^\circ\mathrm{C}$ with $\gamma_f = 6\ \mathrm{J/m^2}$. (b) Fitted toughness distribution for the C(T) specimen tested at $T = -110^\circ\mathrm{C}, -91^\circ\mathrm{C}$ and $-60^\circ\mathrm{C}$ with varying values for the effective fracture energy, γ_f.

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